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Development of Sumitomo High Toughness Process (SHT) for Arctic Grade Line Pipes

DEVELOPMENT OF SUMITOMO HIGH TOUGHNESS PROCESS (SHT)
FOR ARCTIC GRADE LINE PIPES

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Since 500 thousand tons of pipes for Alaska Pipeline were produced, controlled rolling techniques of plates have been advanced in many mills in the world, using ordinary rolling mill lines. Special controlled rolling (SHT), the innovational controlled rolling process of line pipe steel for Arctic uses, has been developed in Sumitomo, equipping a special furnace in [illegible]ng mill line. By this process, heavy plates (32 mm) with sufficient crack arrestability even at -60 C can be produced by ordinary HSLA steels.

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Development of Sumitomo High Toughness Process (SHT) for Arctic Grade Line Pipes

EIJI MIYOSHI **TAKAHIDE TANAKA** **NORIHIKO NOZAKI** **MINORU FUKUDA**

INTRODUCTION

The essential properties of high test line pipe steels are strength, toughness, good field weldability, mass-productivity, and economy. For arctic grade line pipes, low temperature toughness is especially required. The SHT process, which is a kind of the controlled rolling, has been developed to fulfill the aforementioned purpose. The SHT process remarkably improves the low temperature toughness of HSLA steels without any loss of strength and it also keeps high mill productivity.

The SHT process consists of two steps of rolling (1)[1] as is shown in Fig. 1. At the first step, slabs are heated at conventional heating temperature, then rolled to the certain intermediate thickness and are cooled to the temperature lower than the critical transformation temperature. At the second step, steels are reheated at the normalizing temperature and are controlled rolled to the specified plate thickness. The SHT process is schematically shown in Fig. 2. The rolling path of SHT process is shown in Fig. 2 by solid line, compared with that of conventional or ordinary controlled rolling indicated by the dotted line. The temperature in the special SHT reheating furnace is precisely controlled by computer. These SHT facilities were established in Kashima Works of Sumitomo Metal Industries Ltd. in 1975. Over 10,000 tons of line pipes for extremely low temperature service have been produced.

The production yields of ordinary controlled rolling are said to decrease to about 70 percent of that of conventional rolling. The productivity of the SHT process is larger than 95 percent of the conventional one, because the SHT rolling chance can be arranged most effectively among conventional rolling chances.

The advantage of the SHT process is to obtain the low temperature (-50 C) toughness even in heavier thickness of pipes (e.g., over 32 mm) without significant amount of nickel addition. However, 1 percent nickel is added to obtain toughness at heat affected zone of weld joints for the pipes to be used at -60 C. Consequently, the SHT is an economical process to produce the arctic grade line pipes.

PROPERTIES OF PIPES BY THE SHT PROCESS

From over 10,000 tons of pipes, three typical examples — (a) X-70 line pipes with 18.3-mm wall thickness, (b) heavy wall X-65 with 26.3- and 31.8-mm wall thickness, and (c) heavy wall products for extra low temperature service at -60 C — are shown in this report. Properties of SHT pipe are compared to those of the ordinary controlled rolled and the quenched and tempered pipes. All these plates were rolled at Kashima Works and formed to pipes at Kashima Works or Wakayama Works.

Table 1 shows the dimensions and chemical compositions of the tested samples in this report. Table 2 indicates the results of tensile and toughness properties. Figs. 3 and 4 show the transition curves of transverse Charpy absorbed energy. In these figures, pipes made by the SHT and the ordinary controlled rolling processes are compared for the same composition of steels. The transition temperature of the SHT pipes are lower than that of ordinary controlled rolled pipes. Fig. 5 shows the transition curves of B-DWTT (2, 3). It is seen that brittle fracture arrestability can be maintained at a temperature down to -60 C in the SHT pipes with the same chemical composition as that of the ordinary HSLA steels.

Table 3 shows properties of welded pipes. Ordinary two-pass SAW welding process (one pass from each side) was applied to all pipes. Welding materials were specially developed for arctic

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[1] Numbers in parentheses designate References at end of paper.

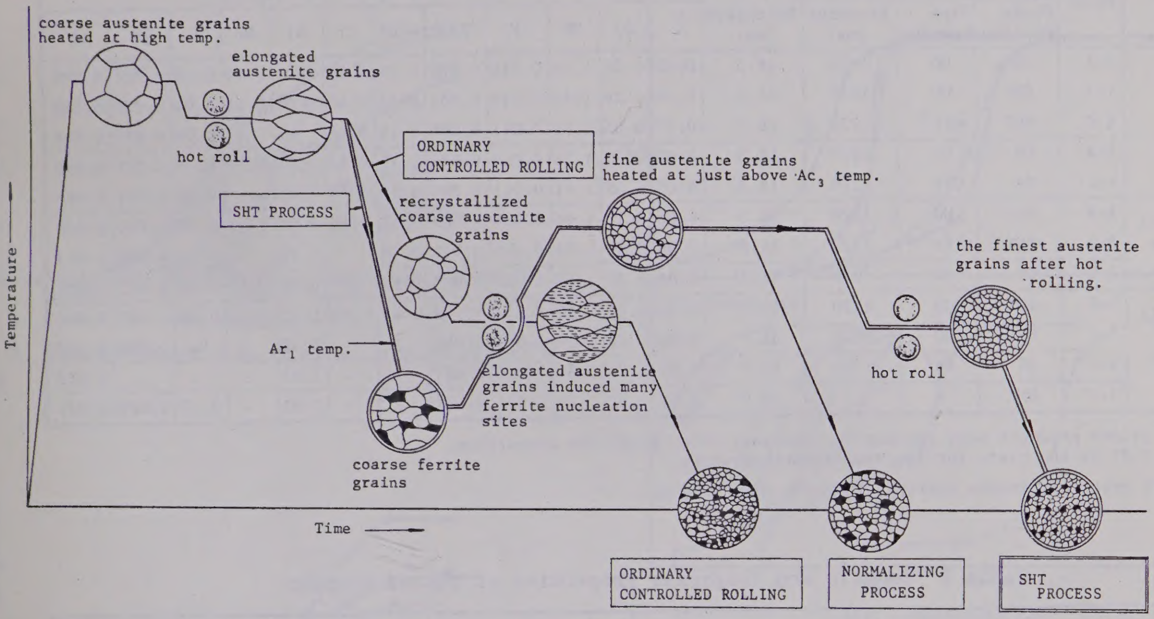

coarse austenite grains heated at high temp.
elongated austenite grains
hot roll
ORDINARY CONTROLLED ROLLING
fine austenite grains heated at just above Ac$_3$ temp.
SHT PROCESS
recrystallized coarse austenite grains
the finest austenite grains after hot rolling.
Temperature
Ar$_1$ temp.
elongated austenite grains induced many ferrite nucleation sites
hot roll
coarse ferrite grains
Time
ORDINARY CONTROLLED ROLLING
NORMALIZING PROCESS
SHT PROCESS

Fig. 1 Schematic diagram of microstructure changes by SHT, ordinary controlled rolling and normalizing method

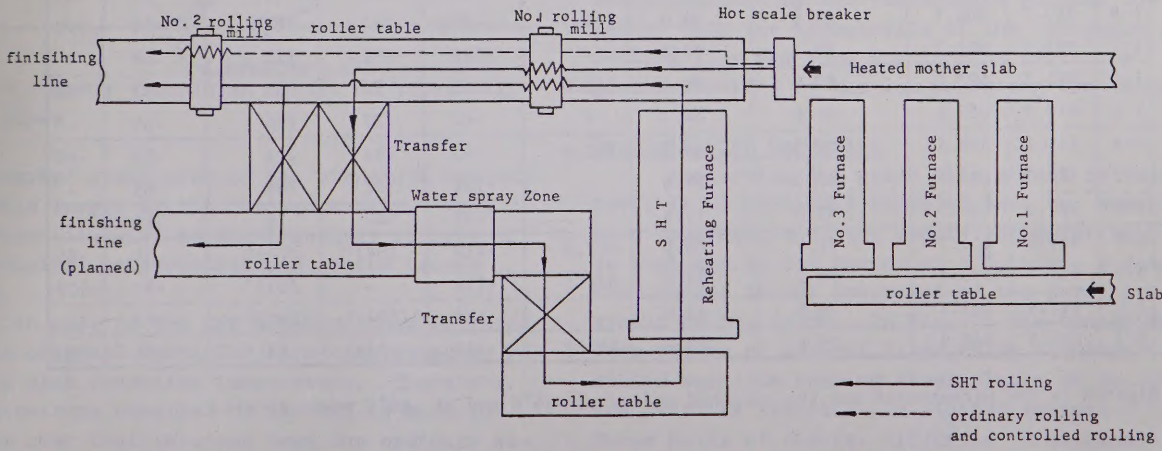


Fig. 2 Layout of rolling units and flow of rolled material

services (4). The location of the notch of Charpy test pieces is selected to indicate the least toughness in the neighborhood of the welded seam of pipes. In the table, heavy wall pipes of 2-B and 2-D containing little or no nickel, respectively, show the low toughness at -40 C, while a pipe of 3-F with 1 percent of nickel shows 67 J even at -60 C. Pipe 3-G, containing 3.5 percent of nickel, shows almost the same toughness at -60 C. Toughness of the pipe body is satisfactorily improved by the SHT process, but to improve the HAZ toughness at a temperature below -40 C, 1 percent of nickel additions is essential.

METALLURGICAL FOUNDATIONS FOR THE EXCELLENT TOUGHNESS

Grain Size

In the following paragraphs, several

Table 1 Dimensions and Chemical Compositions of Supplied Products

Grade	Mark†	Process		Dimensions		Chemical compositions (wt. %)											
		Plate Rolling	Pipe Forming	Diameter (mm)	Thickness (mm)	C	Si	Mn	P	S	Cu	Cr	Ni	Mo	V	Nb	Al
X-70 Line Pipes	1-A	SHT	UO	1220	18.3	0.09	0.28	1.26	0.019	0.005	0.15	0.10	0.01	0.16	0.08	0.027	0.045
	1-B	SHT	UO	1220	18.3	0.06	0.26	1.46	0.016	0.004	0.13	0.10	0.01	0.11	0.08	0.030	0.038
	1-C	SHT	CFE$^{\nabla}$	1220	18.3	0.07	0.35	1.51	0.011	0.004	0.18	0.16	-	-	0.09	0.041	0.046
	1-A'	CR	UO	1220	18.3	0.09	0.28	1.26	0.019	0.005	0.15	0.10	0.01	0.16	0.08	0.027	0.045
	1-C'	CR	CFE	1220	18.3	0.07	0.35	1.51	0.011	0.004	0.18	0.16	-	-	0.09	0.041	0.046
Heavy wall X-65 Line Pipes	2-B	SHT	UO	1220	26.3	0.06	0.26	1.46	0.016	0.004	0.13	0.10	0.01	0.11	0.08	0.030	0.038
	2-D	SHT	RB$^{\nabla}$	1220	31.8	0.09	0.27	1.34	0.009	0.005	0.28	-	0.31	-	0.10	0.028	0.063
	2-E'	CR	UO	920	26.3	0.06	0.31	1.50	0.027	0.003	0.19	-	0.20	0.16	0.06	0.033	0.035
Heavy wall Products for -60°C	3-F	SHT	UO	1220	31.7	0.06	0.26	1.46	0.021	0.008	0.01	0.03	1.13	0.15	0.08	0.027	0.060
	3-G	SHT	UO	1220	31.7	0.06	0.25	1.19	0.007	0.006	-	-	3.42	-	-	0.030	0.036
	3-H'	QT	RB$^{\nabla}$	1220	31.7	0.06	0.26	0.56	0.006	0.007	-	-	3.62	-	-	-	0.042
	3-J*	SHT	-	-	30.0	0.10	0.32	1.38	0.021	0.004	-	-	0.25	-	0.07	0.020	0.029

† : primed products were produced by ordinary CR or by QT for comparison.
3-J* is the plate for low temperature services.

∇ CFE means continuous forming, RB means Roll Bending.

Table 2 Tensile and Toughness Properties of Tested Samples

Grade	Mark	Transverse tensile properties				Through thickness tensile strength (N/mm^2)	Transverse toughness properties				
		Yield stress (N/mm^2)	Tensile strength (N/mm^2)	Yield to tensile ratio (%)	Elongation (%)		2V Charpy test			B-DWTT 85% SATT (°C)	NRL NDT (°C)
							50% FATT (°C)	Absorbed energy (J) at -25°C	Absorbed energy (J) at -45°C		
X-70 Line Pipe	1-A	535.6	652.8	81.4	30.0	605.8	-86	95	84	-57	-65
	1-B	513.7	605.8	84.8	34.0	581.3	-90	227	187	-52	-70
	1-C	493.1	611.7	80.5	35.1	-	-141	152	122	-68	-
	1-A'	581.3	693.0	83.9	29.6	663.6	-68	54	52	-31	-65
	1-C'	541.0	637.1	85.0	34.3	-	-82	222	156	-31	-
Heavy Wall X-65	2-B	455.8	575.4	79.3	44.1	563.6	-93	244	205	-53	-65
	2-D	454.7	547.8	83.0	58.0	-	-101	192	168	-65	-
	2-E'	531.2	634.2	83.8	42.6	-	-47	166	78	-16	-
Thick Products for -60°C	3-F	464.2	611.7	71.7	40.9	617.5	-119	(94)*	(76)*	-64	-85
	3-G	534.2	675.4	79.2	40.0	634.2	-150	-	(81)*	-94	-125
	3-H'	447.9	585.2	76.5	43.5	-	-124	(294)*	(294)*	-73	-
	3-J	435.1	544.9	79.8	42.5	552.7	-113	(84)*	(82)*	-78	-100

* Figures in the parenthesis are the absorbed energy at -45°C and at -60°C respectively.

metallurgical reasons for the excellent toughness obtainable by the SHT are to be illustrated. One of the authors has made clear (5) the γ to α transformation characteristics in the severe controlled rolling process, as shown in Fig. 6(b), which is named the type II transformation. The ultra-fine ferrite grains are yielded at the deformation bands in the unrecrystallized austenite grains just after, or sometimes on, the final low temperature rolling. However, a few large austenite grains are necessarily remained untransformed, and they grow somewhat larger ferrite and pearlite grains during the slow cooling because of the lack of nucleation sites for ferrite within them. Hence, the microstructure of the controlled rolled products inevitably contains a few islands of large ferrites and pearlites, or sometimes bainites. In other words, the microstructure obtained by the severe controlled rolling becomes heterogeneous. Although, the size of grains that were borne at the deformation bands is extremely fine,

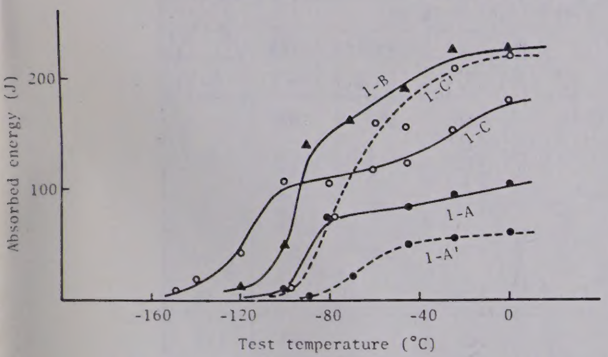


Fig. 3 Charpy transition curves of X-70 line pipes

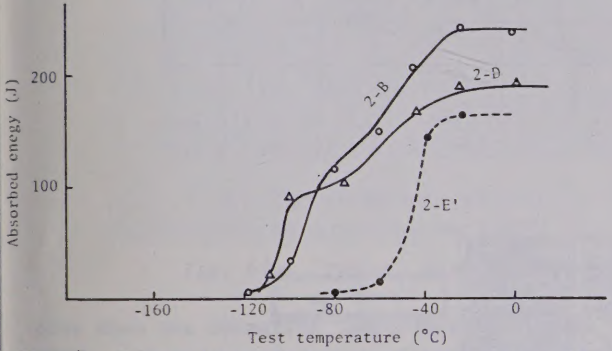


Fig. 4 Charpy transition curves of heavy wall line pipes

the average grain size of the structure becomes a little larger by this heterogeneity. This heterogeneity will be significantly reduced if the original unrecrystallized grains become finer.

In case of the SHT process, this refining of the original austenite is accomplished by reducing slab reheating temperature. Therefore, the structure obtained by the SHT is much improved over that obtained from the ordinary severest controlled rolling, mainly due to reducing the heterogeneity and also to refining the average grain size. Fig. 6(b) shows the typical microstructure of the SHT (right) compared to that of the severely controlled rolling (left). It is recognized that the finest portions of the structure in both steels are quite similar. As interpreted in the foregoing, the structure of the ordinary controlled rolling contains several islands of large grains and these islands disappeared in the SHT plates.

Briefly speaking, the SHT utilizes the

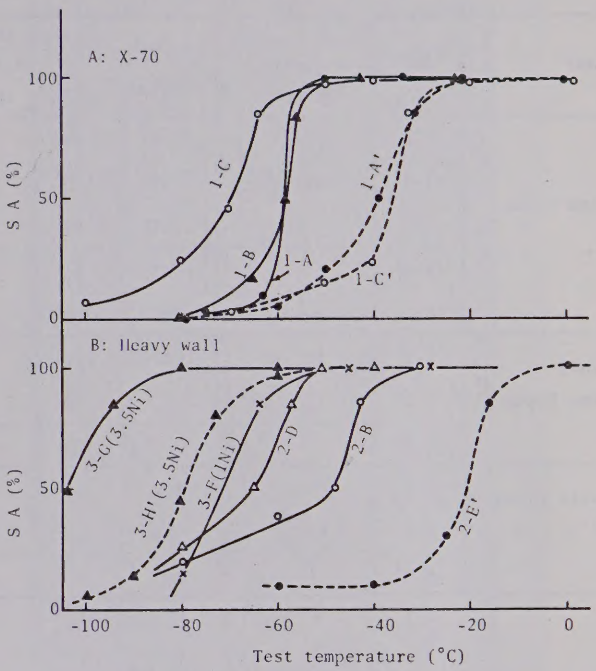


Fig. 5 Transverse B-DWTT transition curves of all tested pipes

ultra-fine grains obtainable by the type II transformation by low temperature rolling together with the homogeneity of the structure resulting from the fine homogeneous austenite grains provided by the low reheating temperature.

Precipitation Hardening

Concerning the grain size effect on toughness, it is generally accepted that the transition temperature of the steels can quantitatively be analyzed by the parameter $d^{-\frac{1}{2}}$, where d denotes the average linear intercept of the ferrite grains by the crack. In Fig. 7, the transition temperature of several 11-mm-thick aluminum killed vanadium bearing steel plates of ferrite and pearlite structure is plotted against $d^{-\frac{1}{2}}$. Three heats of steels, differing in manganese content, are rolled under the different conditions. The open symbols are the results of plates produced by the ordinary processes including normalizing and a few degrees of controlled rolling. They are arranged along a band, the slope of which is somewhat larger than that indicated by Petch (6). The results of the plates made by the SHT are shown by the closed symbols, and it is noticed that they all exist far below the bands of the open symbols.

Hence, although the improvement in toughness by the SHT is attributed, at least partly,

Table 3 Properties of Welded Part of Supplied Pipes

Grade	Mark	Weld tensile strength (N/mm^2)	Charpy absorbed energy of weld (J)							
			H.A.Z.				Deposited metal			
			at 0°C	-25°C	-45°C	-60°C	at 0°C	-25°C	-45°C	-60°C
H-70 Line Pipes	1-A	652.8	108	54	33	-	146	137	96	-
	1-B	621.5	129	103	54	-	177	160	126	-
	1-C	-	155	89	61	34	174	124	46	25
	1-A'	-	113	61	35	-	137	136	107	-
	1-C'	-	162	66	44	26	124	67	46	24
Heavy wall X-65 Line Pipes	2-B	588.1	267	117	33	22	180	160	137	112
	2-D	-	-	110	-	-	-	184	-	-
	2-E'	-	-	103	-	-	-	122	-	-
Heavy wall pipes for -60°C	3-F	621.5	108	-	91	67	-	-	86	92
	3-G	622.4	88	-	54	54	89	-	86	92
	3-H'	-	281	-	-	137	-	-	-	120

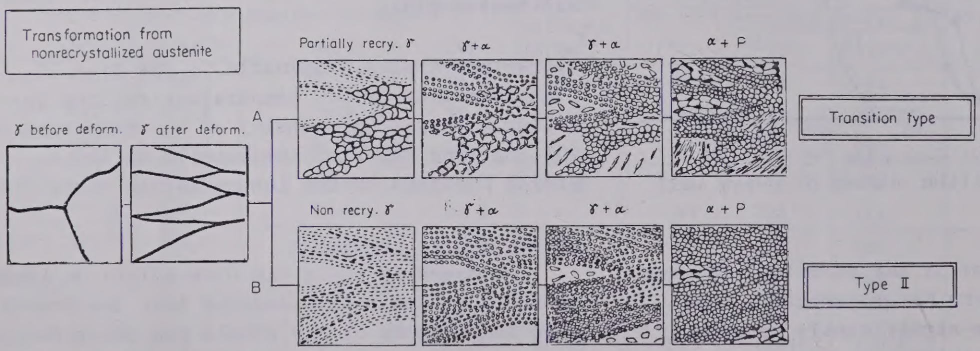
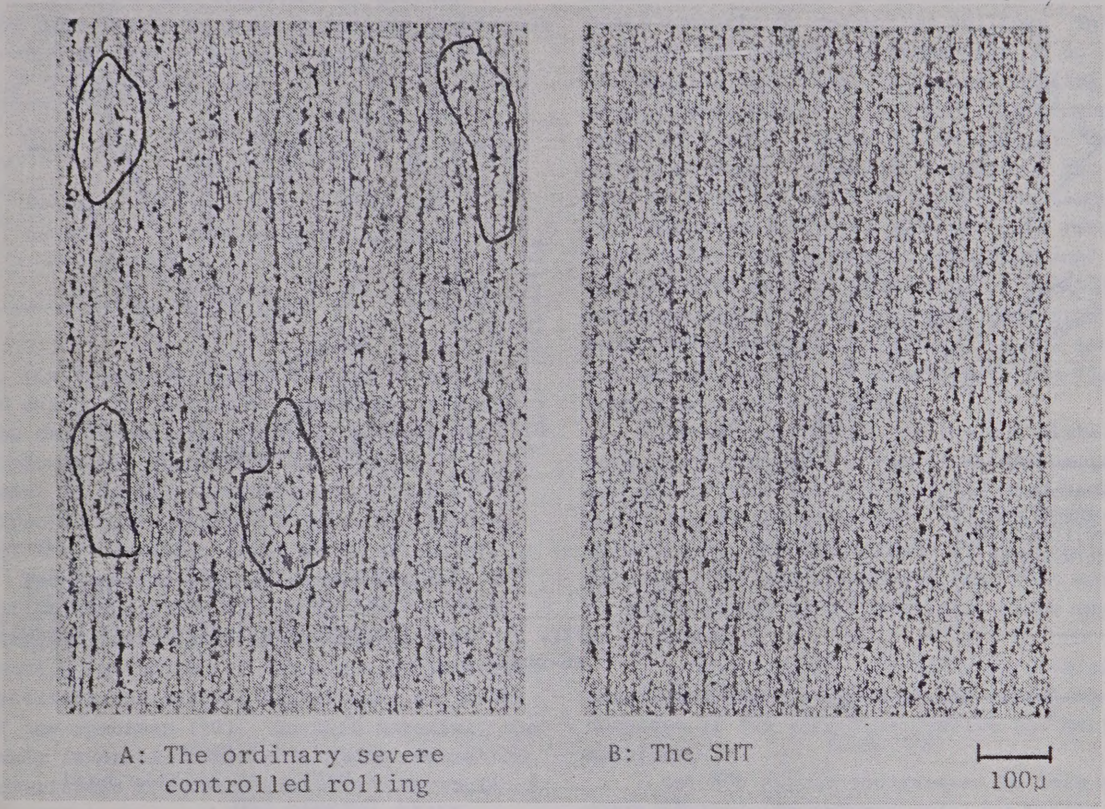


Fig. 6(a) The γ to α transformation mechanism of the severely controlled rolled steels

to the grain refining of the plates described in the previous section, the authors consider that some other metallurgical factors contribute at the same time to the superior toughness. It is formerly proposed by one of the authors (7) that the type of precipitation hardening may change by the severe low temperature rolling.

The effect of precipitation hardening and embrittling by the fixed amount of Nb or V are summarized in Fig. 8. The increase in tensile strength by the addition of V and Nb (ΔTS) is plotted in abscissa, and the accompanying deterioration in Charpy transition temperature (ΔFATT) is in the ordinate. The results of plates without rolling, i.e, heat treatment, are plotted by the open symbols, while the results rolled by 50 percent at temperature between 700 and 800 C are shown by the closed symbols. Arrows from open symbols to closed ones show the effect of low temperature rolling on the precipitation hardening and embrittling. It is ascertained that, by this low temperature rolling, the grain sizes of plain carbon aluminum killed steels are equally refined to those of Nb or V bearing steels, and so, in the first approximation, the effect of grain size cancels each other.

Since the arrows are directed downward and have considerable length, it is concluded that the precipitation embrittling is decreased markedly by the 50 percent low temperature rolling, irrespective to the reheating temperature. Furthermore, the precipitation hardening seems to increase considerably by the rolling in the


A: The ordinary severe controlled rolling

B: The SHT

Fig. 6(b) Typical microstructure of the SHT and of the controlled rolling

case when the reheating temperature is 900 C (SHT). It is supposed that without rolling, the precipitates operate as secondary coherent particles while they operate as dispersion small incoherent particles if rolled at the low temperature. Unfortunately, however, both particles have not been observed yet in the commercial steels.

It appears that the SHT utilizes the full potential of the dispersion hardening and that appreciable strengthening by the addition of V and Nb can be attainable without any deterioration of toughness.

Crystallographic Orientation

Another significant factor contributing to the superior toughness of the SHT plate is the crystallographic orientation (8). Fig. 9 represents the relative intensity of the reflected X-ray from the (100) plane of the crystals parallel to the rolling surface. The SHT plate contains a larger amount of (100) oriented crystals, especially in the mid-thickness of the plate.

As is seen in the literature, it is theoretically expected that the toughness of the plate is improved by the crystallographic texture of (100) plane parallel to the plate surface (8). Moreover, as the authors suggested earlier (9), the stronger the (100) crystallographic texture, the more sensitive the plate becomes to separations or splittings, in the mean time of crack propagation. Since the separations, by their nature, release the stress triaxiality at the crack tip, the toughness of the plate may apparently be further improved.

This crystallographic (100) texture of the SHT plate is not the direct consequence of the low slab reheating temperature, however. Rather, it is the result of the large amount of low temperature rolling. Therefore, the intensity of the texture can be adjustable preferably at some expense of toughness.

Thus, the extreme fine grains, the change of the character of precipitation hardening from the embrittling type to the tender type, and the favorable crystallographic orientations are the main reasons for the excellent toughness of the SHT plates.

DISCUSSIONS

According to the metallurgical common

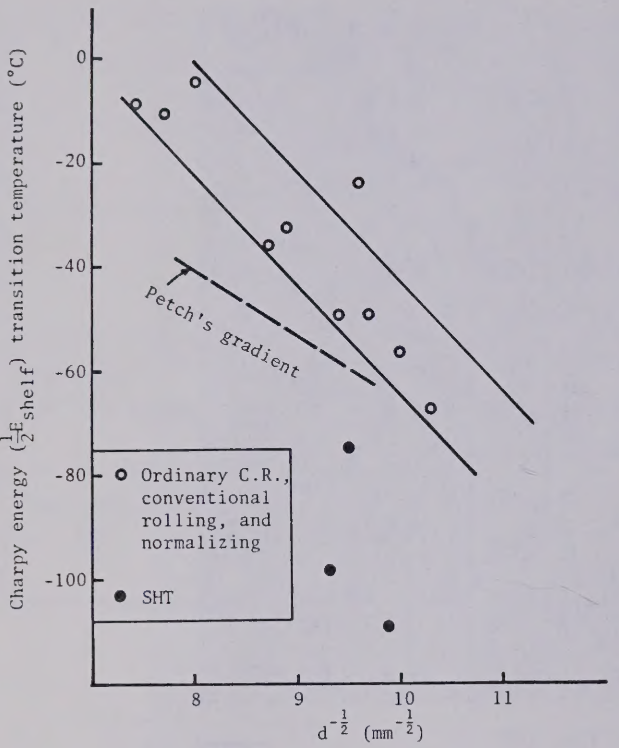
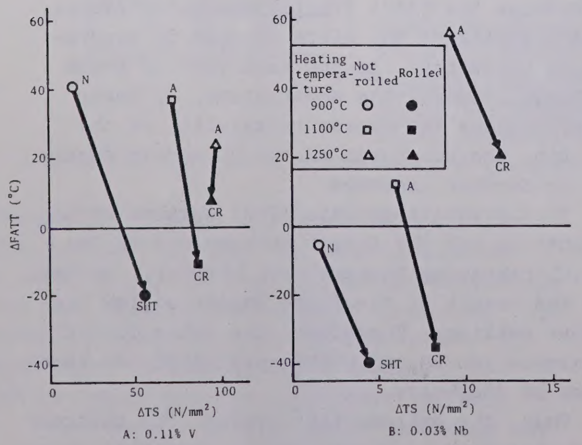


Fig. 7 The plot of temperature versus $d^{-\frac{1}{2}}$ for laboratory melted vanadium bearing steels. (C: 0.17 percent; Si: 0.29 percent; Mn: 0.6∿1.4 percent; V: 0.06 percent; Al: 0.03 percent)



A: 0.11% V

B: 0.03% Nb

Fig. 8 The effect of 0.11 percent V (left) and 0.03 percent Nb (right) on the strength and toughness of Al-killed steel plates. (C: 0.10 percent; Si: 0.30 percent; Mn: 1.32 percent; Al: 0.03 percent)

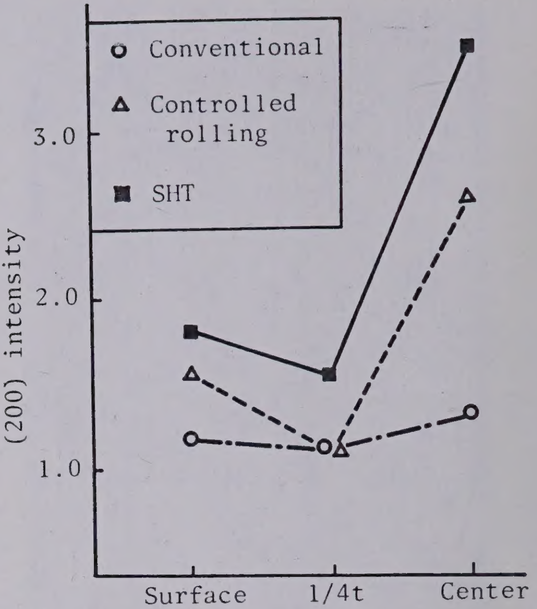


Fig. 9 Crystallographic orientation of as-rolled Nb-steel plates

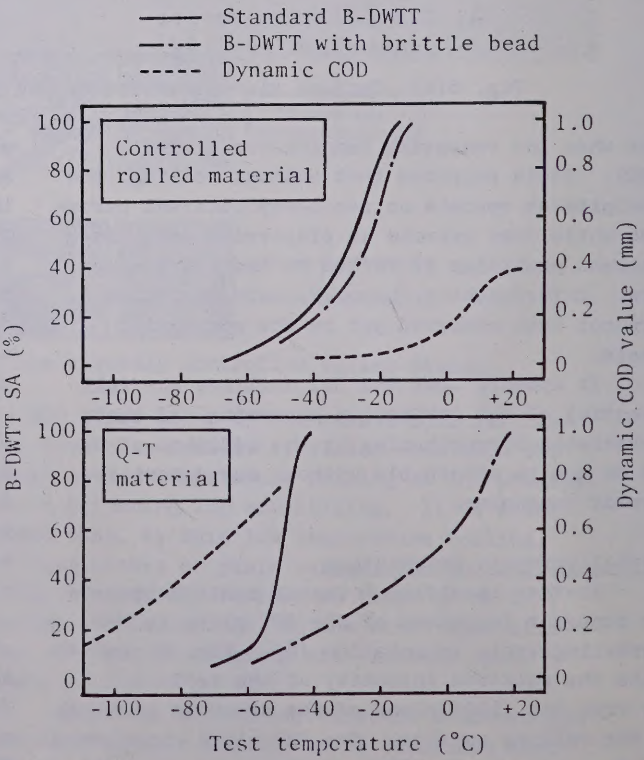


Fig. 10 The initiation and the propagation arrest characteristics of the as-rolled and the Q-T materials

Table 4 Strength and Toughness of Acicular Ferrite Type Steels in Several Rolling Conditions

<table>
<tr><td rowspan="2">Chemical composition</td><td>C</td><td>Si</td><td>Mn</td><td>P</td><td>S</td><td>Mo</td><td>Nb</td><td>Al</td></tr>
<tr><td>0.05</td><td>0.26</td><td>2.46</td><td>0.017</td><td>0.006</td><td>0.24</td><td>0.06</td><td>0.026</td></tr>
<tr><td rowspan="6">Transverse mechanical properties</td><td rowspan="2" colspan="2">Rolling condition</td><td colspan="4">Tensile property (8.5ø)</td><td colspan="2">Charpy test</td></tr>
<tr><td>Yield stress (N/mm^2)</td><td>Tensile strength (N/mm^2)</td><td colspan="2">Elongation (%)</td><td>50% FATT (°C)</td><td>Absorbed energy at 0°C (J)</td></tr>
<tr><td colspan="2">Conventional</td><td>530.2</td><td>750.7</td><td colspan="2">20.5</td><td>+100</td><td>12</td></tr>
<tr><td colspan="2">Controlled</td><td>557.1</td><td>774.0</td><td colspan="2">20.3</td><td>-88</td><td>96</td></tr>
<tr><td colspan="2">Severely controlled</td><td>557.1</td><td>775.2</td><td colspan="2">20.4</td><td>-104</td><td>86</td></tr>
<tr><td colspan="2">SHT</td><td>488.8</td><td>742.4</td><td colspan="2">22.5</td><td>-132</td><td>97</td></tr>
</table>

sense, the toughness of pipes seems to be improved easier by the quench and temper process than by the controlled rolling. However, in practice, it has been found out that the improvement in brittle fracture arrestability of the pipes is quite different from the improvement of the transition temperature of Charpy test. For example, Fig. 10 shows that the transition temperature of B-DWTT of a quenched and tempered material is appreciably raised if the ductile initiation of the crack is prevented by depositing the brittle bead around the notch tip of the specimen (10). In this material, the extremely favorable brittle fracture initiation characteristics are ascertained by the result of dynamic COD test.

Because the crack arrestability should be evaluated to determine whether the brittle crack is arrested or not when a running brittle crack is introduced into the material, this Q-T material is judged to have poor arrestability in spite of excellent transition temperature of B-DWTT and the Charpy test. Contrary to this, the as-rolled materials have generally poor brittle fracture initiation characteristics, and the value of the B-DWTT and the Charpy test gives rather reasonable crack arrestability as shown in the upper part of Fig. 10. Generally speaking, the tempered martensite has relatively superior initiation characteristics and poorer arrestability than the ferrite and pearlite if the apparent toughness value by a customary small size test is equal. However, the high tensile strength over 600 N/mm^2 can be reached more easily by Q-T process.

The authors are now attempting to make such high strength steels with superior arrestability by the SHT or by the tempering after SHT rolling. The microstructure produced is a very fine mixture of ferrite and martensite, like the micro-duplex structure (11).

Recently, the acicular ferrite type steel was developed as a very tough high strength steel (12). The plates of this steel are produced commonly by controlled rolling. Table 4 presents the mechanical properties of an acicular ferrite type steel in several rolling conditions. When rolled by the SHT, the toughness value is further improved with a slight decrease in the strength. However, increasing the magnitude of rolling reduction at the low temperature with simultaneously reducing the slab reheating temperature, the extremely fine poligonal ferrite yielded and the proportion of acicular ferrite decreases. It can be mentioned that the superior property of this steel is not inherently due to the acicular ferrite but to the fine and homogeneous microstructure.

The microstructure of the SHT plate of this steel contained a very fine dispersion of martensite like the micro-duplex structure. For the micro-duplex type high strength steel produced by the SHT, the higher carbon content is tolerable for high toughness, and some addition of vanadium may be preferable in place of a part of Mo and Nb.

In any respect, the SHT will develop in the future the new class of high strength steels composed of the very fine ferrite and martensite mixtures.

CONCLUSIONS

The authors devised and developed the special controlled rolling method named SHT, which is composed of two steps of rolling with an intermediate cooling and reheating.

The proper layout of the mills and the special attachments, including a furnace, was designed without any loss of productivity even when a very low temperature rolling was conducted.

By this new rolling method, the brittle fracture arrestability can be guaranteed down to -40 C in the X-70 grade line pipes, heavy wall pipes with thickness over 25 mm, and heavy plates, without any nickel addition. For the product to be used at temperature below -40 C, the toughness of the 1 percent Ni steel rolled by the SHT is nearly equal to that of the 3.5 percent Ni steel made by the ordinary process (Q-T). The SHT is especially useful if the brittle fracture arrestability of thick products is required.

The excellent toughness of the SHT is due to the extreme fine grains, the tender unembrittling precipitation hardening, and the favorable crystallographic orientations.

The detailed discussions on the crack arrestability of the steel leads us to conclude that the SHT is the easier way to get brittle fracture arrestability than the quench and tempering process.

The SHT is a promising process to mass-produce steel plates for low temperature services.

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